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Size effects in micro-tensile testing of high purity polycrystalline nickel

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ABSTRACT

Tensile properties and fracture characteristics of high purity polycrystalline nickel were investigated using micro-tensile specimens. The material response to the applied load was found to be sensitive to both geometry and sample size. The post-tests examination of the fracture surfaces revealed that the mechanisms leading to failure were associated with the nucleation, growth and coalescence of voids. These mechanisms were studied using finite element implementation of the Gurson–Tvergaard–Needleman (GTN) model. A physical meaning of the model parameters was proposed and validated against experimental data. The model provided good predictions of the failure mode, but did not capture the variability observed for the micro-tensile specimens. The factors such as machining process, surface roughness, and local variations in the microstructure were most likely responsible for these differences.

1. Introduction

Much of the current research and technology development require miniaturization of engineering components. These components are often made of advanced materials, and are expected to have enhanced mechanical properties. In many cases, however, the available quantity of such material is insufficient to measure its mechanical properties by conventional testing procedures. This attracts interest in the use of small-scale mechanical testing instrumentation and specimens having sizes and geometries deviating from the ASTM standards.

The characterization of mechanical properties with small size specimens, however, might not necessarily provide the same mechanical response to loading as conventional counterparts. The mechanism of plastic flow, defects accumulation and ultimate failure might be affected, for example, by the specimen dimensions (Keller, Hug, & Feaugas, 2011; Matic, Kirby, & Jolles, 1988; Sergueeva, Zhou, Meacham, & Branagan, 2009; Zhao et al., 2008), anisotropy and heterogeneity of the material (Alexander & Beyerlein, 2005; Haouaoui, Karaman, & Maier, 2006; Li, Winther, & Hansen, 2006; Lopes, Barlat, Gracio, Ferreira Duarte, & Rauch, 2003), or a small number of grains in the cross-section of the specimen (Keller et al., 2011; Lim, Kim, Lee, Kim, & Kim, 2008; Yang & Lu, 2013). This is particularly true for metals with high ductility, which can possibly generate a larger scatter in the experimental results (e.g. failure data) for the same specimen geometry. One can relate these variations to the fact that there is no specimen free from defects, which can be either intrinsic (microstructural inhomogeneities)

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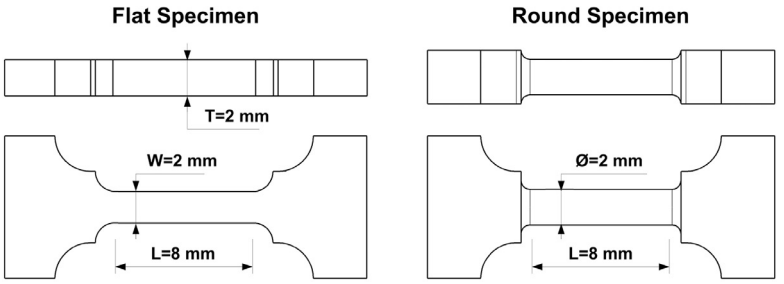

No.	Flat Specimen	
	W × T	L
1	2 × 2	8
2	3 × 3	12
3	4 × 4	16

No.	Round Specimen	
	Ø	L
1	2	8
2	3	12
3	4	16

Fig. 1. Nomenclature and dimensions of the specimens (units: mm).

or extrinsic (surface imperfections). Since the onset of plastic instability in ductile materials occurs at strains which are large compared to elastic strains, the probability of instability from pre-existing defects at the same strains is likely to be low. Some other limitations of small-scale mechanical testing instrumentation are related to the strain measurement of the specimens. Note that the measurements by means of extensometers are generally problematic for the small-size specimens, whereas non-contact measurement techniques are potential sources of the error.

The underlying mechanisms of ductile failure are associated with the nucleation, growth and coalescence of voids. The nucleation of voids is usually initiated through the cracking of inclusions or debonding of inclusion-matrix interfaces. Once this process started, nucleated voids grow and coalesce readily with progressive plastic deformation (Benzerga & Leblond, 2010; Garrison & Moody, 1987; Puttick, 1959). The first attempt to describe the growth of cylindrical and spherical voids in ductile materials dates back to the 1960s (McClintock, 1968; Rice & Tracey, 1969). To this day, a significant consideration in a continuum-mechanics framework has attracted the model developed by Gurson (1977). The author proposed a yield criterion as a function of the void volume fraction (f) for rigid-perfectly plastic solids containing a cumulative volume of spherical voids. This f parameter describes the damage as a ratio between the volume of voids and the volume of the matrix material. In the following years, several improvements and extensions were made to the model, where the most important being introduced by Tvergaard (1981, 1982). Tvergaard proposed two additional factors, q_1 and q_2, in the yield criterion to improve the accuracy of the model prediction. However, a set of generalized and well-fitted parameters has not been found, and many following studies brought them up for discussion (Brocks, Sun, & Hönig, 1995; Faleskog, Gao, & Shih, 1998; Gao, Faleskog & Shih, 1998; Kim, Gao, & Srivatsan, 2004; Koplik & Needleman, 1988). Further improvements were made by Tvergaard and Needleman (1984) with the introduction of the critical void volume fraction (f_c) to provide a criterion for the void coalescence. As discussed in the literature (Kim et al., 2004; Koplik & Needleman, 1988; Zhang, Thaulow, & Ødegård, 2000), the f_c parameter is a material variable and depends on the initial porosity and the total volume fraction of voids prior to coalescence.

Therefore, most of the studies have suggested the Gurson–Tvergaard–Needleman (GTN) model parameters that give a good approximation of the macroscopic mechanical behavior, but which are not linked to the material inherent length scale. In this study, we attempt to define physics-based parameters of the GTN model. This is illustrated with an example of high purity polycrystalline Ni investigated under tensile loading conditions, and with a use of small-size specimens.

2. Experimental procedure

A high purity (99.97 wt.% min.) electrolytic Ni (ERAMET) plate of 10 mm thickness was used as a testing material. The microstructure was characterized in the length–width plane of the plate (referred as the X_1–X_2 plane) using a Zeiss Supra 40VP SEM equipped with a fully automated electron backscatter diffraction (EBSD) analysis system. Prior to the EBSD analysis, the sample was polished with P4000 grit SiC paper and electropolished to reveal the grain structure. The resulting maps were analyzed with the OIM software from TexSem Laboratories (TSL) to determine the average grain size, microtexture and the distribution of grain boundary misorientation. A step size of 0.35 µm was used for the EBSD acquisition. Baseline microhardness testing was performed using a Vickers microhardness tester with a diamond indenter of a square base and an angle of 136° between opposite faces. The tests were performed under the load of 0.1 kgf (dwell time of 7s) at room temperature. Reported microhardness values are an average of at least five readings.

Fig. 1 shows the nomenclature and nominal dimensions of the specimens used in the micro-tensile and conventional tensile tests. The gauge length of the specimens was equal to four times the nominal diameter, similar to the ASTM standard (ASTM E8/E8M). All specimens were cut from a plate with respect to the X_1 direction. The square tensile specimens were cut using spark erosion, whereas the round tensile specimens were machined using a high precision lathe machine. The micro-tensile tests were performed on a tensile/compression stage adapted for in-situ SEM experiments (made in LSPM CNRS-UPR3407 laboratory). The load cell was capable to measure force up to 10 kN in a speed control mode. The stage had a linear scale with a resistive extensometer for high-accuracy non-contact elongation measurements. Therefore, the engineering stress was calculated as the force divided by the initial cross sectional area, while the engineering strain as the current displacement measured by the extensometer divided by the gauge length of the specimen. The engineering stress–